

ALLOY STEEL AND ARTICLE MADE THEREFROM.

FIELD OF THE INVENTION

This invention relates to alloy steels and more particularly to a relatively low cost high strength high toughness Copper-Nickel-Chromium (Cu/Ni/Cr) alloy steel and an article made therefrom.

BACKGROUND OF THE INVENTION

AF1410 (US Patent No 4,076,525) and AirMet 100® (US Patent No 5,087,415) are exemplary of alloy steels (Group A herein), sometimes referred to as superalloys, which combine very high strength and high toughness. One drawback of these alloy steels is their very high cost, because their superior mechanical properties are achieved through the use of scarce and costly alloying elements such as cobalt, nickel and molybdenum, and complex processing, namely double vacuum remelting, aging, refrigeration (Table 3). Their high prices and poor machinability limit their applications to mainly military and aerospace products.

US Patent Nos. 3,713,905 and US Patent No. 6,187,261 are exemplary of high strength steels such as grades AISI 4340 and 300M moderately priced alloy steels (Group B herein). They use traditional chemistries to provide high ultimate and yield strengths after thermal processing, but lack the high toughness of the first (Group A) described alloys.

Typical lower priced martensitic alloy steels (Group C herein) as described in US Patent No. 5,180,450 provide good toughness but lack the high ultimate and yield strengths of the Group A and B alloy steels.

The use of the Group A ultra high strength and high toughness alloy steels is limited because of their prices. Engineers must select from a large numbers of steel alloys to meet either

strength or toughness design requirements and cost objectives. This practice causes high steel making and inventory costs and a need for large amounts of scarce high cost alloying elements. Moderately priced steels combining high strength and high toughness (as Group A) with good machinability and heat treatment properties (as Group B) would have substantial economic value and advance the art of alloy steels.

SUMMARY OF THE INVENTION

The present invention is a relatively low cost high strength high toughness medium Carbon Copper-Nickel- Chromium alloy steel. A primary object of the invention is to provide a lower cost alternative to current high cost Group A Alloy Steels. A still further object of the invention, in addition to the forgoing object is to provide a competitive cost alternative with significantly higher toughness than the Group B Alloy Steels. A still further object, in addition to the foregoing objects, is to reduce the use of scarce alloying elements such as Cobalt. A still further object in addition to the foregoing objects is to expand the use of high strength high toughness alloy steels.

In a first embodiment of the invention, a medium carbon lower cost alternative is provided to the high cost Grade A alloy steels comprising by weight percent about 0.40-1.00 % Copper, about 2.50-8.00% Nickel, and about 0.80-3.50 % Chromium.

In a second embodiment of the invention, a high toughness alternative is provided to the Grade B alloy steels comprising by weight percent about 0.55-0.70 % Copper, about 1.0-6.0 % Nickel, and about 1.0-2.2% Chromium.

Additional features and benefits will be apparent by reference to the ensuing detailed descriptions of preferred embodiments which discloses the best mode contemplated in carrying

out the invention. The property in which exclusive rights are claimed is set forth in the numbered claims following the detailed description of the preferred embodiment.

DETAILED DESCRIPTION OF THE INVENTION

5 The present invention is the result of a development of cost effective alloy steels, realized through considering the effects of alloying elements and their costs on strength and toughness properties of alloy steels. Selections of suitable alloying elements and their concentrations took into account the known atomic structures of the elements; the solid solubilities of the elements in alpha-iron; the capabilities of the elements to modify the electron structures of solid solutions in
10 the alpha-iron lattice, and the adsorption activities of alloying elements with respect to the interfaces of metal systems. Another important consideration was the ability of developmental alloys to respond effectively to surface strengthening by carburizing and nitriding.

 Table 1 is a summary of alloys tested, their chemical composition (alloys 1-20) and mechanical properties (alloys 1-18).

15 Table 2 shows properties of alloy 19 subjected to carburizing and alloy 20 subjected to nitriding.

 Table 3 shows comparison of mechanical properties and costs of alloying elements of known Group A and Group B target alloys, and several alloys according to present invention (Examples 1, 5, 8 from Table 1).

 Specific ratios of Silicon/ Copper (Si/ Cu), Carbon/ Vanadium (C/ V) and C+V/ Nickel (Ni) were applied to modify and control the diffusion mechanisms of interstitial elements (C, N, O, H) for reducing the probability of brittle phases forming on the grain and phase boundaries during tempering and for modifying and controlling the kinetics of phase and structural

transformations during heating and cooling.

The high adsorption activity of Copper in alpha-iron caused Copper atoms to quickly migrate toward grain and phase boundaries during tempering. Copper together with Silicon formed solid solution zones with low ion potentials, precluding carbide and nitride phases from forming in boundary areas. Copper also improved mechanical properties such as yield to tensile strength ratios, ductility, impact resistance, machinability and corrosion resistance. Copper was determined to be best used in an alloy steel in a ratio of Si to Cu of approximately 1.0 - 2.5.

Concentrations of copper over 1.0%wt. Cu were detrimental to an alloy's toughness because of the formation of epsilon-phase (solid solution of Fe in Cu) on the boundaries of martensite crystals after quenching and tempering. Relatively high amounts of copper in alloy steels as exemplified by US Patent No. 3,713,905 results in relatively low toughness.

Nickel (Ni) in the present invention is the primary contributor to high toughness. Nickel increases the electron density of the alpha-iron based solid solution, transforming alpha-iron lattice into a symmetrical and therefore, a more ductile alpha-iron lattice due to a reduced sharing of covalent bonds. While additions of Ni within a claimed range, for any given level of Carbon within the claimed range, had relatively little effect on the alloy's hardness and ultimate strength after quenching and tempering, the yield strength significantly decreased. At the low end of a claimed carbon range (about 0.22 - 0.30%) adding Ni in an amount over 3.5-4.0% was counterproductive in terms of an alloy's cost, since same yield and toughness properties could be achieved by simply reducing the level of C in the alloy.

At the high end of a claimed carbon range (0.50 - 0.55% C), nickel in amounts over 3.5-4.0% adversely affected both ultimate and yield strength. Up to 8.0% Ni can be used in the mid-carbon range of this alloy (0.30 - 0.45% C) to obtain high toughness and, preferably, up to 6.0%

Ni to obtain optimum combination of strength and toughness.

Chromium (Cr) increased strength, hardenability and temper resistance. In the preferred composition when alloys were subjected to nitriding or carbo-nitriding, Cr was responsible for forming high hardness zones of fine dispersed CrN clusters, coherent with alpha-iron matrix. When the chromium content in the steel exceeded 3.5% wt., alloy's ductile properties were adversely affected in the tempering temperature range of 180°C-650°C.

Silicon (Si) - in the selected ranges strengthened the steel matrix by increasing the share of covalent bonding between atoms in a solid solution. The strengthening effect is also due to its effect on the dislocating formations at intercrystalline and phase boundaries, leading to polygonization of Si-rich areas which provides additional channels for diffusion of carbon during tempering, carburizing and nitriding.

It was determined that the positive effect of Silicon on mechanical properties of the subject steel starts at a silicon concentration of 0.5% wt. Silicon concentrations of over 1.5% wt. caused undesirable carbide and nitride phases to form in the grain boundaries. Silicon in the interval of 0.5 to 1.5% wt. in a steel alloy contributed to improved hardenability and temper resistance, and reduced the transformation temperature from the viscous to the brittle state. Silicon also produces some negative effects on the properties of steel, for example, by causing significant grain growth during long heat cycles, such as annealing. These effects are suppressed in the invention by copper and vanadium.

Vanadium (V) affected the structure and properties of the steel in several ways. First, by forming in austenite fine dispersed particles of carbides, nitrides and carbo-nitrides, which controlled the size and shape of grains and then, consequently, the size of the ferrite grains after "gamma-alpha" transformations. This in turn, affected the strength and ductility of the steel.

Second, by precipitating vanadium based, fine dispersed secondary carbides and carbo-nitrides during quenching and tempering, the vanadium deterred re-crystallization of ferrite and promoted formation in the ferrite of a very fine equi-axial grain. Third, by influencing the kinetics and morphology of the “gamma-alpha prime” transformation products during quenching and tempering. By forming fine dispersed carbides and carbo-nitrides in the austenite grain body, vanadium promoted formation of fine massive martensite, which provided the alloy with good combination of strength and ductility properties after quenching.

Titanium (Ti) in the present compositions acts in similar way as vanadium. Our studies showed that alloy cost can be further reduced by substituting Ti, despite its poorer than vanadium solubility in iron. Ti is a more active carbide and nitride forming element than vanadium, being at the same time less expensive as pure metal and ferroalloy.

The medium carbon (C) embodiment required contents in excess of 0.2% C to meet strength requirements. The principal strengthening mechanism was the tempering of the carbon martensites, which produced a precipitation of carbide particles generally detrimental to high toughness. As carbon alone was increased there was an increased tendency for microcracking due to increased lattice strains present as a result of higher tetragonal distortion. This condition was alleviated by adding substantial amounts of solid strengtheners, e.g. Ni, Mn which reduced the level of carbon necessary to attain high strength. The strength / toughness properties were best balanced in the range of 0.35 - 0.45 %wt C.

Manganese (Mn) in the present composition is somewhat higher than is usual in steels of this type. Mn was necessary as a deoxidizer in the steel making process and, in addition, it is a strong contributor to hardenability. If less than 0.50 %wt. was present, the effects of elimination of hot shortness due to manganese sulfide formation and improvement in machinability was not

be attained to the necessary degree. If above 1.50 %wt Mn. was present, it promoted retention of austenite and the strength of the alloy was adversely affected.

Sulfur (S) improved machinability by facilitating chip breakage. However sulfur above 0.045%wt S was detrimental to physical properties such as ductility and impact strength. Despite these drawbacks however sulfur served the useful function of improving machinability.

Phosphorus (P) - also aided machinability by facilitating chip breakage. However phosphorus above 0.025 was detrimental to physical properties such as ductility and impact strength following mid and high tempering (660-1200°F). Strength and toughness of low tempered martensite (300-400°F) were little influenced by P content of up to 0.15%wt.

Sample alloys were produced in an open air 100lb induction furnace and cast into cylindrical 4.0" diameter graphite molds. Liquid metal was poured from 2920+/-20°F. 50 lb ingots were subjected to homogenizing annealing at 1725°F for 9 hours; then cooled within the furnace to 935°F, then cooled in air to room temperature. Thereafter ingots were heated to 2110°F and rolled or forged to a final size of approximately 1" diameter. Finished bars were further annealed for 9 hrs at 1210°F, Standard ASTM tensile bars and impact V-notch (Charpy) test samples were machined and heat treated. Alloy compositions, quenching and tempering temperatures, and mechanical properties are shown in Table 1.

An optimum combination of strength and toughness was obtained with alloys containing 0.35-0.45% carbon and 2.5 - 6.0% nickel, after quenching from 1580-1625°F (after 1 hour soak time) in a salt bath (or oil quenched, or air cooled) and subsequent tempering at 350 - 400°F for 5-6 hours. Subsequent refrigeration somewhat increased the level of yield strength (Table 1).

Alloys according to the invention having compositions within the ranges of: carbon 0.22-0.32 wt.%, copper 0.40-1.00 wt.%, chromium 0.80-1.25 wt.%, nickel 1.00-3.50 wt.%, can be

effectively strengthened (case hardened) by carburizing and nitro-carburizing.

Tensile and Charpy V-notch samples from the present alloy comprising by weight 0.28% C, 0.93% Mn, 1.04% Cr, 0.97% Si, 0.56% Cu, 3.25% Ni and 0.35% V were carburized for 9 hrs at 1724°F. Samples were normalized at 1625°F and tempered for 6 hours at 1210°F, then quenched from 1544°F in oil and tempered for 6 hours at 350°F. The mechanical properties obtained from these samples are presented in Table 2.

Several alloy samples according to the invention of the same composition were through-carburized to a carbon concentration of 0.65 wt.% in the sample cross-section and tested for fracture toughness with results presented in Table 2.

Alloys according to the invention in the range of concentrations: carbon 0.33-0.55 wt.%, copper 0.40-1.00 wt.%, chromium 1.20-3.50 wt.%, nickel 1.00-3.50 wt.%, can be effectively strengthened by low temperature nitriding.

Tensile and Charpy V-notch samples from an alloy comprising by weight 0.42% C, 0.87% Mn, 2.04% Cr, 1.02% Si, 0.55% Cu, 2.13% Ni and 0.35% V were quenched from 1625°F and tempered for 6 hours at 1040°F, then nitrided in ion nitriding furnace for 24 hrs at 932°F. Mechanical properties obtained from these samples are presented in Table 2.

Although only several embodiments have been described, it is not our intention to limit the scope of our invention to the disclosed embodiments since other embodiments can be derived from what is claimed by changes that are obvious to persons skilled in the art without departing from spirit thereof.

TABLE 1

EXAMPLE No	CHEMICAL COMPOSITION							MECHANICAL PROPERTIES					
	Ni	C	Mn	Si	Cr	Cu	V	HRC	UTS ksi	YS ksi	Elong. %	R.A. %	KCV ft-lbs
1	1.64	0.37	0.91	0.94	1.11	0.55	0.37	54	290	225	12.0	37.0	29
2	2.46	0.45	0.96	1.13	1.31	0.53	0.26	54	301	238	13.8	47.8	31
3	2.46	0.51	1.07	1.26	1.2	0.53	0.25	55	328	232	11.2	27.4	23
4	2.48	0.42	1.14	1.23	1.23	0.53	0.25	55	301	235	12.0	31.2	31
5	2.52	0.4	1.17	1.19	1.21	0.53	0.25	53	297	232	12.6	33.9	33
6	3.74	0.38	0.83	1.01	2.02	0.52	0.34	53	278	211	14.3	40.3	25
7	4.13	0.36	0.86	0.89	1.08	0.54	0.37	52	279	219	12.5	42.0	36
8	4.29	0.39	0.84	0.76	1.25	0.68	0.25	54	287	242	14.2	47.9	30
9	4.42	0.42	1.19	1.31	1.21	0.53	0.25	53	298	226	13.8	41.0	31
10	4.43	0.37	0.76	1.06	1.2	0.53	0.25	52	272	219	14.8	54.1	42
11	4.44	0.51	0.96	1.13	1.15	0.52	0.25	54	313	218	13.7	44.2	28
12	4.47	0.45	0.88	1.05	1.14	0.52	0.25	52	300	218	13.9	44.4	30
13	5.47	0.38	0.81	0.93	1.86	0.53	0.93	50	256	200	13.3	41.8	26
14	5.96	0.38	0.65	0.78	1.22	0.67	0.24	53	281	235	14.5	49.8	32
15	6.84	0.39	1.21	1.44	1.21	0.52	0.25	50	270	190	15.0	39.0	34
16	6.84	0.43	1.18	1.3	1.21	0.53	0.24	51	289	198	16.2	42.3	29
17	6.91	0.46	0.88	0.96	1.17	0.52	0.23	51	291	189	15.1	40.2	24
18	6.97	0.51	0.68	0.88	1.1	0.53	0.24	52	303	175	14.1	31.7	23
19	3.25	0.28	0.93	0.97	1.04	0.56	0.35						
20	2.13	0.42	0.87	1.02	2.04	0.55	0.35						

TABLE 2

EXAMPLE No	TERMO- CHEMICAL TREATMENT	CASE DEPTH IN (MM)	HARDNESS		UTS ksi	YS ksi	Elong %	R.A. %	KCV ft-lbs	Fracture Toughness ksi√in
			Surface	Core						
19	Carburizing: 1724 F/9 Hrs Quenching: 1545°F Temper: 350°F / 5 Hrs	0.07 (1.8)	HRC 59	HRC 45	241	232	8.4	26.9	13.2	
	Carburizing: 1724 F/60 Hrs Quenching: 1545°F Temper: 350°F / 5 Hrs	Through	HRC 59	HRC 59						48, 52, 54
20	Quenching: 1625°F Temper: 1040°F / 3 Hrs Ion Nitriding: 935°F / 24Hrs	0.022 (0.55)	HV 1202ksi	HRC 42	198	185	11	29.4	17.6	

TABLE 3

	STEEL GRADE	CHEMICAL COMPOSITION	HEAT TREATMENT	MECHANICAL CHARACTERISTICS						CHARGE MATERIAL COST \$ / Lb *
				HRC	UTS ksi	YS ksi	Elong. %	R.A. %	KCV ft-lbs	
Group "A"	AF 1410	C-0.19; Cr-2.00; Ni-10.0; Co-14.0; Mo-1.00	Quench. 1625°F Refrig. -100°F Aged 900°F	52	254	224	16	69	48	1.83
	AirMet 100	C-0.25; Cr-3.10; Ni-10.0; Co-13.5; Mo-1.20	Quench. 1625°F Refrig. -100°F Aged 900°F	54	285	250	14	65	30	1.82
	Example 8 Table 1	C-0.39; Mn-0.84; Si-0.76; Cr-1.25; Ni-4.29; Cu-0.68 V-0.25	Quench. 1625°F Temper. 350°F	54	287	242	14.2	47.9	30	0.21
Group "B"	AISI 4340	C-0.43; Mn-0.80; Si-0.30; Cr-0.90; Ni-2.0; Mo-0.30	Quench. 1580°F Temper. 400°F	53	270	225	12	40	21	0.13
	300M	C-0.40; Mn-0.75; Si-1.60; Cr-0.85; Ni-1.85; Mo-0.4 V-0.10	Quench. 1685°F Temper. 400°F	54	289	245	9	34	22	0.15
	US Patent No 3,713,905 Example 5	C-0.40; Mn-0.21; Si-0.98; Cr-1.07; Ni-3.04; Mo-0.71 Cu-2.00; V-0.12	Quench. 1525°F Temper. 300°F	54.5	309	207	8.7	27.8	11	0.24
	Example 1 Table 1	C-0.37; Mn-0.91; Si-0.94; Cr-1.11; Ni-1.64; Cu-0.55; V-0.37	Quench. 1625°F Temper. 350°F	54	290	225	12	37	29	0.11
	Example 5 Table 1	C-0.40; Mn-1.17; Si-1.19; Cr-1.21; Ni-2.52; Cu-0.53; V-0.25	Quench. 1625°F Temper. 350°F	53	297	232	12.6	33.9	33	0.14

" * " - approximate charge cost for each alloy based on average 2002-2003 LME prices